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Fatigue of Copper-Zinc Alloys at 100°K

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FOREWORD

This report was prepared by the Creep and Dynamics Section, Metals and Ceramics Laboratory, Directorate of Materials and Processes, Deputy for Technology, Aeronautical Systems Division. The research was initiated under Project No. 7351, "Metallic Materials", Task No. 735106, "Behavior of Metals", with Mr. J. A. Roberson acting as project engineer.

The work was conducted during the period from January 1962 to October 1962.

The assistance of Messrs. K. D. Shimmin and J. R. Schmermund in the design and construction of the test apparatus and that of TSgt. J. M. Stewart in conducting the experiments is gratefully acknowledged.

ABSTRACT

Fatigue tests were conducted to determine the S-N diagrams for a series of copper-zinc alloys at 100°K. An attempt was made to relate the endurance limit behavior to both stacking fault energy and yield strength, but no simple relationship was found. It is suggested that high stacking fault energy increases the cyclic work hardening rate by increasing the probability of dislocation intersection and jog formation. The increase in work hardening rate is reflected in a decrease in plastic strain amplitude, and a subsequent increase in fatigue life. These arguments are bounded on one side by considerations of the yield strength of the alloys, and on the other side by consideration of the dominant mechanisms operative in short and long life fatigue.

This technical documentary report has been reviewed and is approved.

W. J. Trapp

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INTRODUCTION

A number of observers (refs 1-4) have commented on the importance of dislocation cross-slip in the fatigue behavior of f.c.c. metals. This mechanism can contribute to either fatigue hardening or fatigue softening depending on the strain amplitude and the initial condition of the material. In either case, cross-slip is commonly believed to provide a mechanism for dynamic recovery by allowing screw dislocations to escape from piled up groups and tangles, and it may lead to substructure formation. This is rather significant since the fatigue crack seems to propagate along the subgrain boundaries. Since the difficulty of producing cross-slip is dependent on the stacking fault energy (γ) of the material, other conditions being held constant, this parameter ought to provide a basis on which some fatigue phenomena can be explained.

Since the stacking fault energy of copper-zinc alloys varies in a known manner with zinc content, this alloy system was chosen for investigation. To minimize the effects of diffusion, either to dislocations (Suzuki locking) or to the fatigue crack itself, the experiment was performed near liquid nitrogen temperatures. Even so, it was not found possible to isolate the effects of stacking fault energy completely. The variation in yield strength with composition represents the largest single complicating factor and has prevented us from establishing a unique relationship between stacking fault energy and fatigue behavior.

BACKGROUND

A dislocation is the boundary between a slipped and a non-slipped region in a crystal. In the simplest case, the pure edge dislocation, it can be imagined as the edge of an extra half-plane of atoms somewhere in the lattice. One of the most fundamental characteristics of a dislocation is its Burger's vector. It is this characteristic which determines the directions which the dislocation can move without mass transport. The edge dislocation has a Burger's vector which is everywhere perpendicular to it. These two perpendicular line components determine a plane, the slip plane, and the dislocation can not leave this plane without diffusion of vacancies or interstitials to or from the dislocation line. The screw dislocation has a Burger's vector which is everywhere parallel to it. These two line components do not determine any particular plane, so this type of dislocation should be able to move freely in any direction.

In f.c.c. metals it is energetically favorable for these dislocations to be split into partial dislocations. Two partial dislocations are connected by a strip of stacking fault. The width of the stacking fault is governed, under ideal conditions, by the stacking fault energy. In order for an extended screw dislocation to slip from one plane to another (cross-slip), it is first necessary for it to collapse into perfect screw orientation. It is not essential for the entire length of the dislocation to take part in the cross-slip process.

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A stacking fault is simply a deviation from normal stacking order. The close packed planes in a f.c.c. lattice can be identified ABCABC, depending on the relative positions of the atoms as successive layers of lattice planes are built up. The sequence in h.c.p. lattices is ABABAB. The simplest way to visualize a stacking fault in the f.c.c. lattice is to remove one plane, then rejoin the lattice without lateral motion. This will result in a sequence ABCABABC, which is equivalent to introducing a very thin layer of h.c.p. metal in the f.c.c. lattice.

It would be expected that the presence of a stacking fault would have an effect on the mechanical behavior of a crystal, and this is indeed observed without difficulty in single crystals. The stress strain curve of a single crystal exhibits a region of easy glide, one of linear work hardening, and a final, dynamic recovery stage where cross-slip becomes profuse. The resolved shear stress required to promote profuse cross-slip is called τ_{III}, and some authorities have taken this as an indirect measure of the stacking fault energy, that is, it is the stress necessary to collapse the partial dislocation - stacking fault ribbon arrangement prior to cross-slip (ref 5).

Stacking fault energy can be considered to be analogous to surface tension for the purposes of this discussion. Metals with high values of stacking fault energy have narrow extended dislocations, and cross-slip in these metals is easily produced. Metals with low values of stacking fault energy have wide extended dislocations, and cross-slip in these metals is difficult. The collapse of these dislocations can also be assisted by thermal fluctuations, and it is common to find a strong dependence of τ_{III} on testing temperature. No satisfactory theoretical explanation of stacking fault energy has been developed, nor has it yet been established whether it is dependent on temperature.

The stress necessary to produce cross-slip, then, is dependent on testing temperature and stacking fault energy. (It also depends on the magnitude of the internal stresses exerted on the stacking fault ribbon by other dislocations.) One observer (ref 3) has reported that the temperature dependence of the endurance limit for a life of 10^6 cycles in copper single crystals is the same as the temperature dependence of τ_{III}. This relation was observed only at very low temperatures (below about 160°K). Several observers (refs 1-4) have reported a general trend of increasing endurance limit (at long lives) with decreasing values of stacking fault energy. If one accepts the polygonization and subsequent crack growth concept mentioned in the Introduction, it is easy to develop a model which would predict such behavior in properly oriented single crystals or in polycrystals at very low stresses. At higher stresses the requirements of constancy of volume and grain boundary continuity cause the model to fail for polycrystalline materials.

EXPERIMENTAL PROCEDURE

The fatigue tests were conducted in an insulated chamber held near liquid nitrogen temperature. The design of the chamber was to some extent dictated by the testing machine, which was a 300 Kg Schenck Pulsator. The chamber was fitted to the bed of the testing machine, and the pull bars were inserted through the ends of the chamber.

To conserve liquid nitrogen and avoid the problems of containing a liquid, it was decided to run the tests at 100°K, a temperature slightly above the boiling point of N_2.

After the specimen had been measured and installed, two copper-constantan thermocouples were tied to it, and the chamber was filled with cotton. Liquid nitrogen was admitted to the chamber so that the specimen temperature was controlled by the rate of evaporation of the nitrogen through the cotton. One thermocouple signal was fed into an on-off type of controller, which actuated a 3-way solenoid value, thus increasing or releasing the pressure on the liquid nitrogen container. The second thermocouple signal was fed into a portable millivolt recorder, and the reference junction was immersed in liquid nitrogen. During calibration it was found that there was no perceptible temperature difference between a thermocouple which was tied to the specimen and one which was soldered to it. It was also found that the specimen grips were cooled to the same temperature as the specimen.

It was necessary to install a deflector to shield the specimen from direct contact with the liquid nitrogen stream to prevent sudden temperature variations. This restriction did not apply to the grips since they were quite massive in comparison. The outside ends of the pull bars were water warmed so that no calibration errors were introduced into the stress monitoring system. Two hours were allowed for cooling the system and establishing thermal and dimensional equilibrium before the load was applied to the specimen. At low stress levels (life greater than about 10^4 cycles), the temperature was easily controlled at $100° \pm 2°K$. At higher stress levels, the heat generated in the specimen caused larger temperature variations.

The frequency of the testing machine was 3600 cycles per minute, and all specimens were tested to failure. The stress level was adjusted prior to starting the machine, and it was not possible to correct it after starting. For this reason a few of the tests were run with a mean stress slightly different from zero. Other than this, the tests were run in reversed axial tension-compression. Stress was determined by microscopic observation of the deflection of the calibrated spring which serves as part of the loading system.

The specimens were annealed and electropolished prior to testing. They were therefore quite free from scratches and machining stresses. Some geometrical imperfections were introduced during electropolishing, and it was necessary to discard a few of the specimens. Because of the heating problem it was not possible to determine the upper end of the S-N diagram; however, that end of the diagram was not considered to be important for the purposes of this investigation. Most of the tests were run at a stress which would produce failure at between 10^4 and 10^6 cycles. It would have been desirable to run more tests at lives greater than 10^6 cycles, but the rate of nitrogen consumption (about 2 liters per hour) made this impractical. The same procedure was followed in installing, cooling, and loading all of the specimens so that any possible errors would be consistent and would not affect the relative behavior of the different alloys.

The specimens used for the tensile tests were machined from the same lots of material, and were given the same heat treatment as the fatigue specimens (annealed in Argon for 1 hour at 500°C). Tensile tests were conducted with the sample immersed in a bath of liquid nitrogen. Strain and strain rate were determined from cross head motion, which was constant at 0.1 inch per minute. The active gauge length was one and one-half inches. The physical and mechanical properties of the alloys are shown in table 1.

TABLE I

PHYSICAL AND MECHANICAL PROPERTIES OF Cu - Zn ALLOYS

Composition (Wt. %)	93.4 - 6.6	90 - 10	85 - 15	70 - 30
Annealed Grain Size (Microns)	23	22	20	30
0.2 % Yield Stress at 77.4 °K (psi)	13,650	15,250	19,450	21,850
UTS at 77.4 °K (psi)	54,000	57,400	65,800	68,400
Stacking Fault Energy at 300 °K (erg/cm^2) a)	63	36	24	14
τ_{III} at 100° K (psi) b)	8,000	8,200	8,750	12,200

a) Ref. 6

b) Ref. 7

RESULTS AND DISCUSSION

Fatigue data were obtained in the form of standard S-N curves. Figure 1 displays the results for the four alloys tested. The individual curves and the numerical data are presented in the Appendix. The tensile stress strain curves are shown in figure 2. For purposes of later discussion, it is convenient to show the variation of certain parameters as a function of composition. In figure 3, this variation is shown for the 0.2 percent yield stress (σ_y), the stacking fault energy (γ) as measured by Howie and Swann (ref 6), and the endurance limit (σ_e) at 10^6 cycles. In this figure the change in composition is plotted as the number of free electrons per atom, $^e/a$. In the case of copper-zinc alloys the addition of X percent of zinc increases the free electron concentration by approximately X percent.

In figure 4(b) the endurance limit at various lives is shown as a function of γ; no simple relationship is apparent except at lives of 10^6 cycles or greater. In figure 4(a) the endurance limits are plotted against the yield strengths (σ_y); again, no simple relationship is apparent. However, for any given life it can be seen that the two plots show opposite slopes. The reason for this dual and opposite dependence of σ_e on σ_y and γ can be seen in figure 3. Both σ_y and γ are functions of the free electron density in the alloys, and these curves have opposite slopes. From the changes in slopes in figures 4(a) and 4(b) and from the cross over points in figure 1 it is evident that a change in the fatigue behavior occurs between lives of 10^4 and 10^5 cycles.

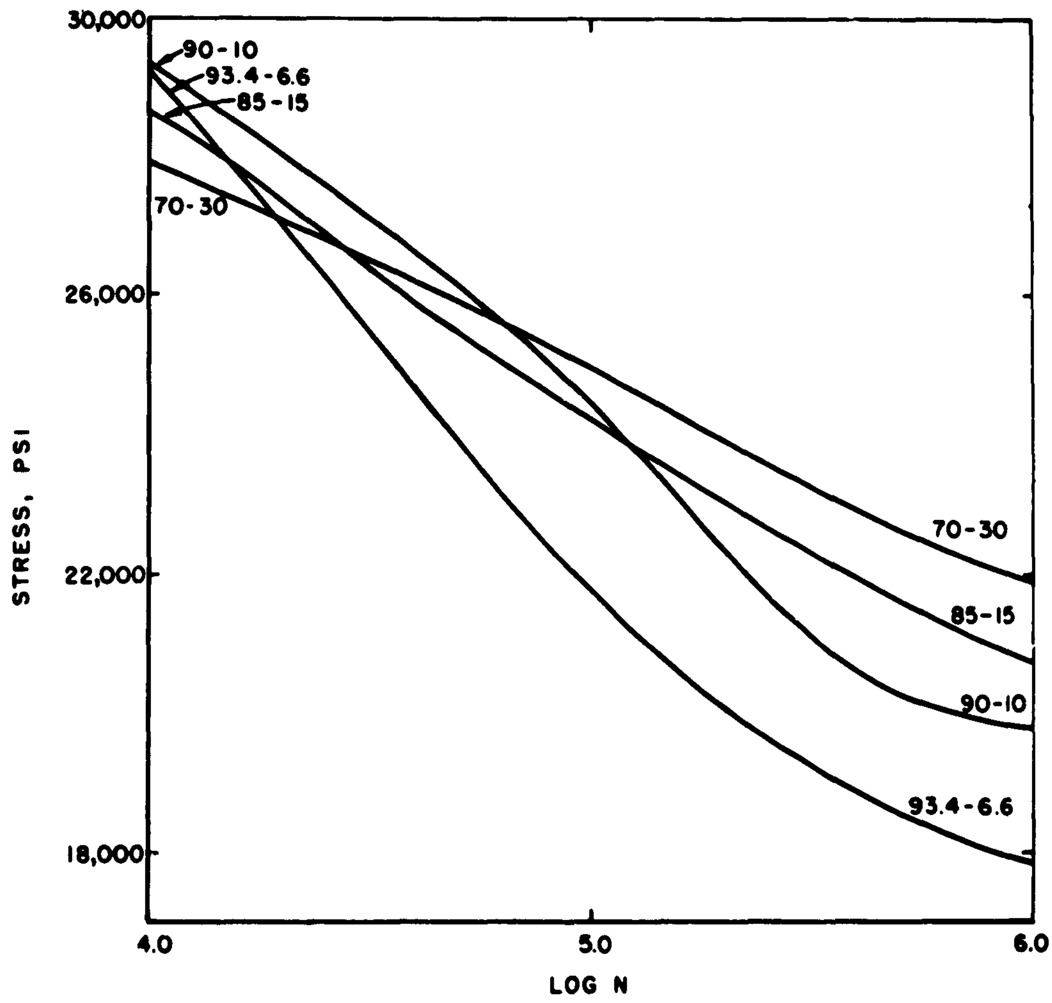

Figure 1. Fatigue Behavior of Copper-Zinc Alloys at 100°K

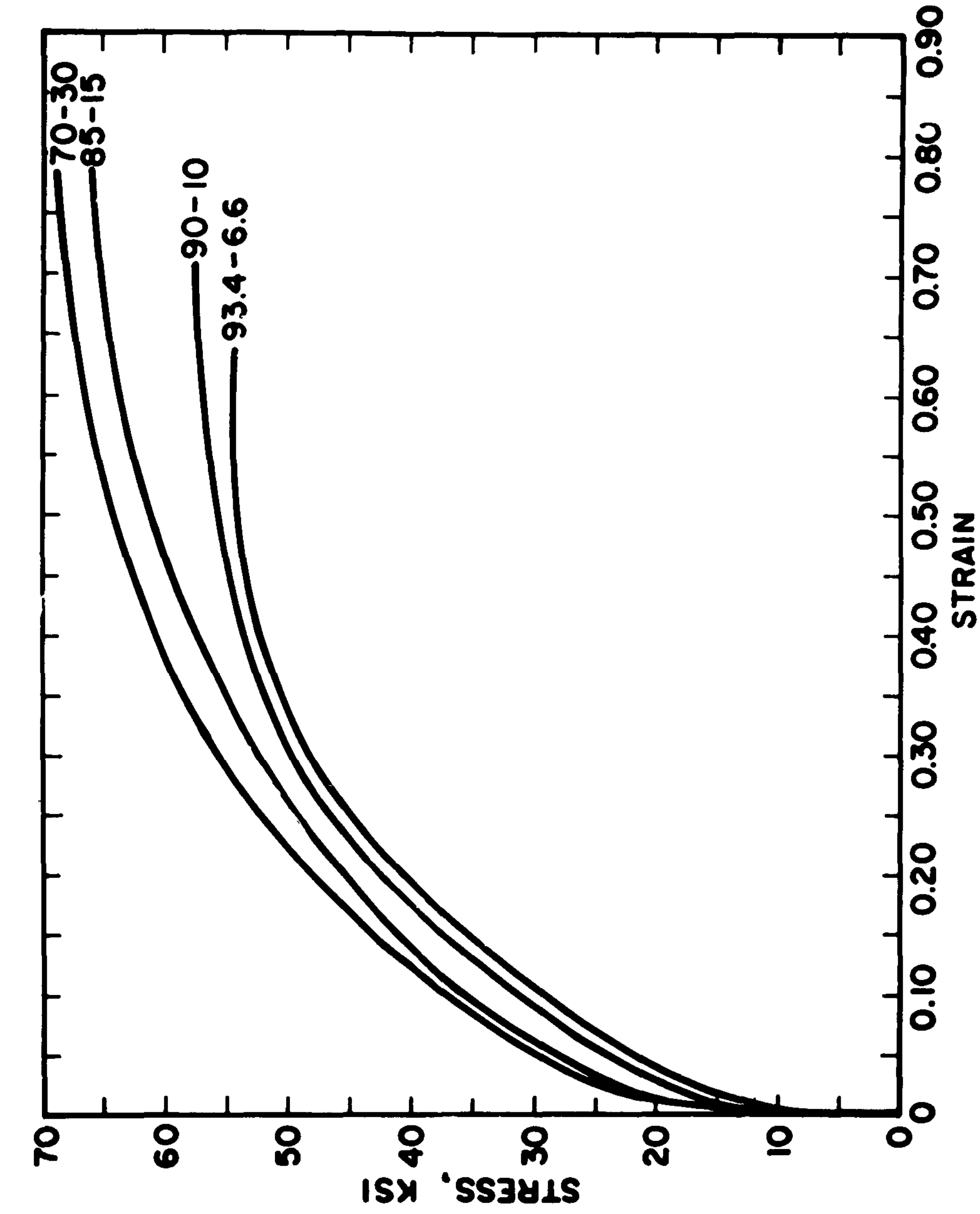

Figure 2. Tensile Behavior of Copper-Zinc Alloys at 77.4°K

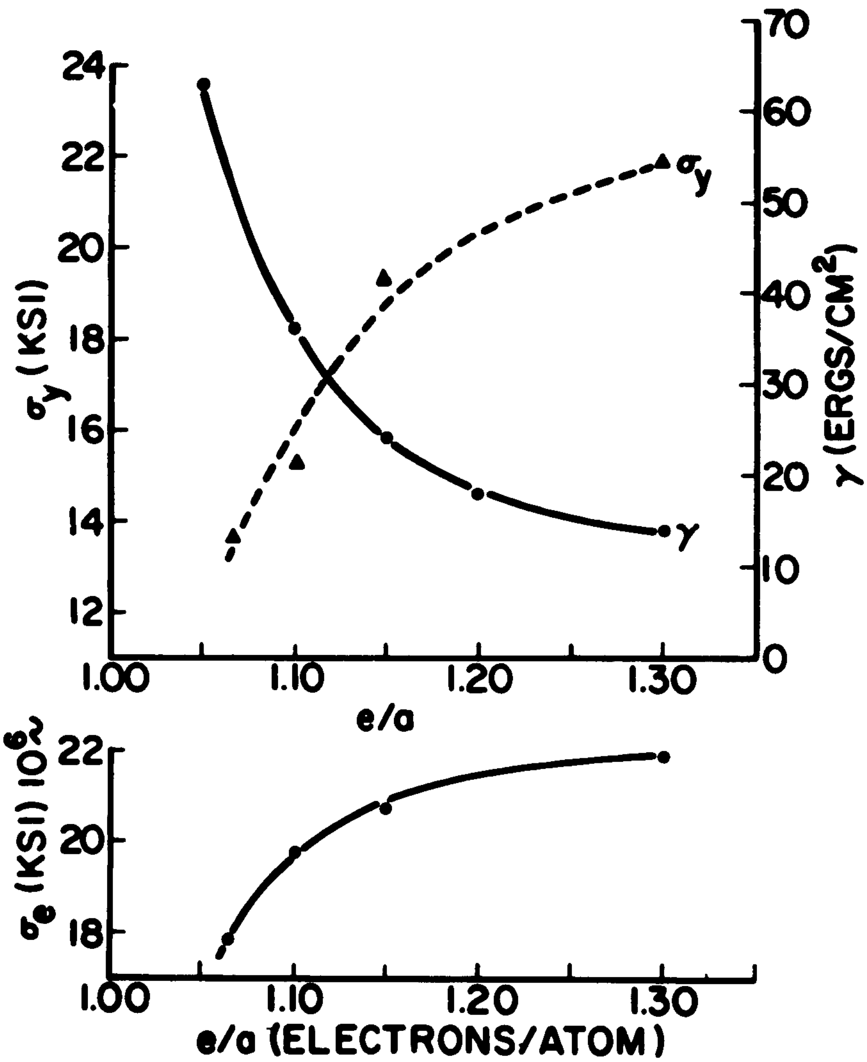

Figure 3. Variation of Properties with Composition

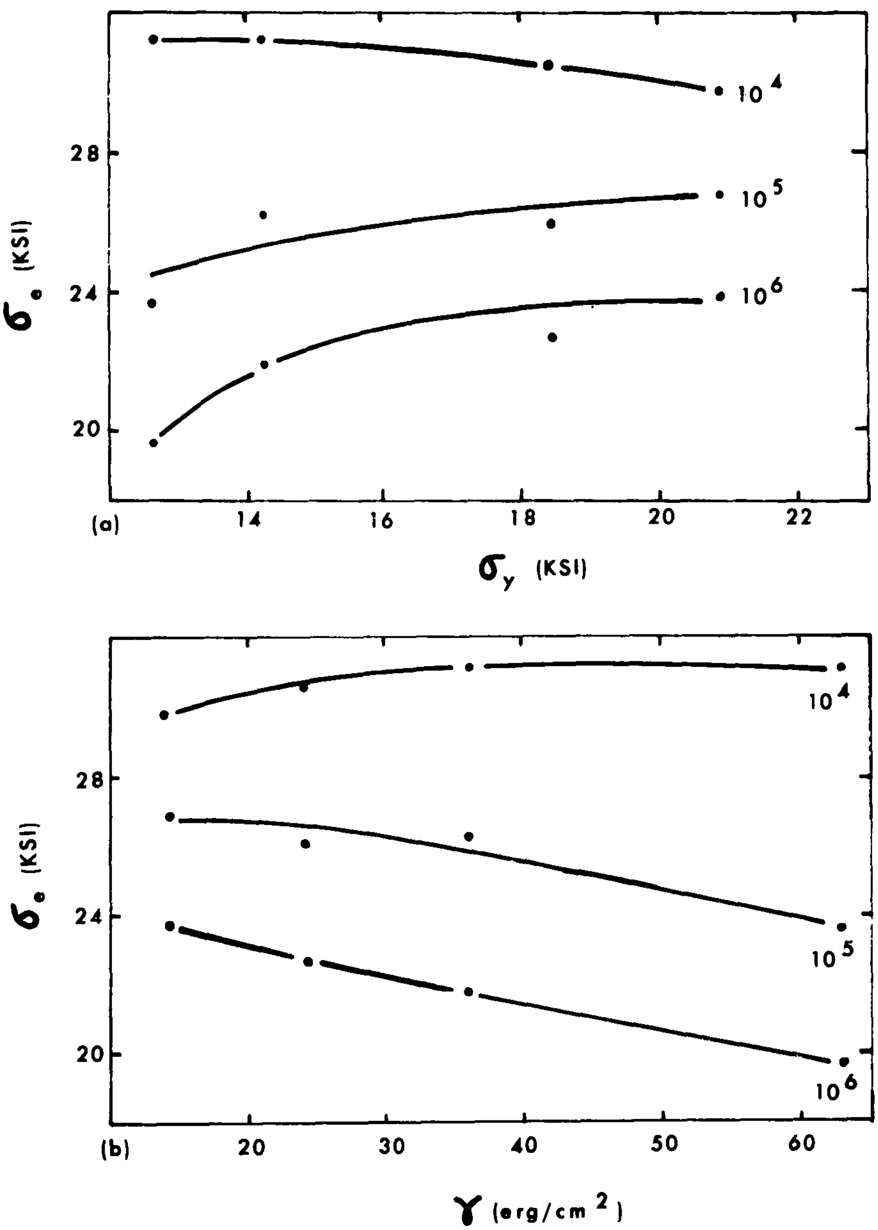

Figure 4. Variation of Endurance Limit with Yield Strength (a) and Stacking Fault Energy (b)

From the arguments which will be developed in the following paragraphs, it would appear that this behavior is a reflection of the variation in relative yield strengths rather than of a fundamental change in the mechanisms of fatigue.

The ideal experiment, in which one tests materials of constant yield strength but varying stacking fault energy, appears almost impossible if one requires further that the grain size be constant. It is well known that for copper-base alloys the yield stress is a unique function of e/a (refs 8 and 9). That is, for a given value of e/a, the yield strength appears to be independent of the solute element (with the exception of tin). Moreover, for the alloys which have been investigated (Cu-Zn, Cu-Ge, and Cu-Al), γ depends inversely on e/a. Thus, by varying e/a (composition), a change in both σ_y and γ is a necessary consequence.

In an attempt to study fatigue behavior while avoiding the complexities caused by variations in the yield strength, we have normalized the S-N curves to the initial yield strength of each alloy. In figure 5 the reduced stress is plotted versus cycles to failure. Two interesting points should be noted. First, the long life dependence on composition has been inverted and the cross over points between 10^4 and 10^5 cycles no longer exist. Second, at long lives the two softer alloys and the two harder alloys merge into single curves.

To understand why a 95-5 alloy will last longer in fatigue than a 70-30 alloy for a given multiple of the yield stress, it is necessary to consider the reduced stress-strain curve, figure 6. Notice that the low zinc alloys (high γ) possess a much higher work hardening rate and hence can sustain a larger multiple of the yield stress before failure.

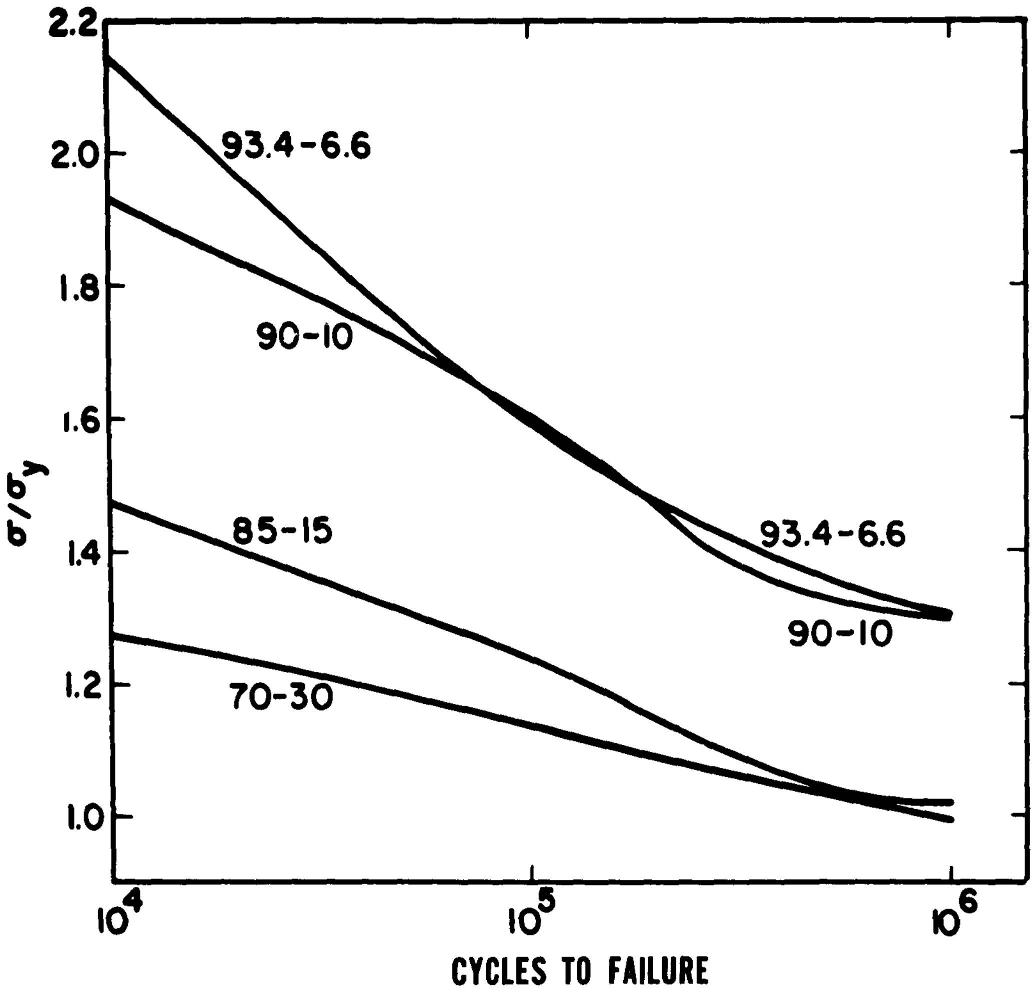

Figure 5. Fatigue Behavior Normalized to Yield Strength

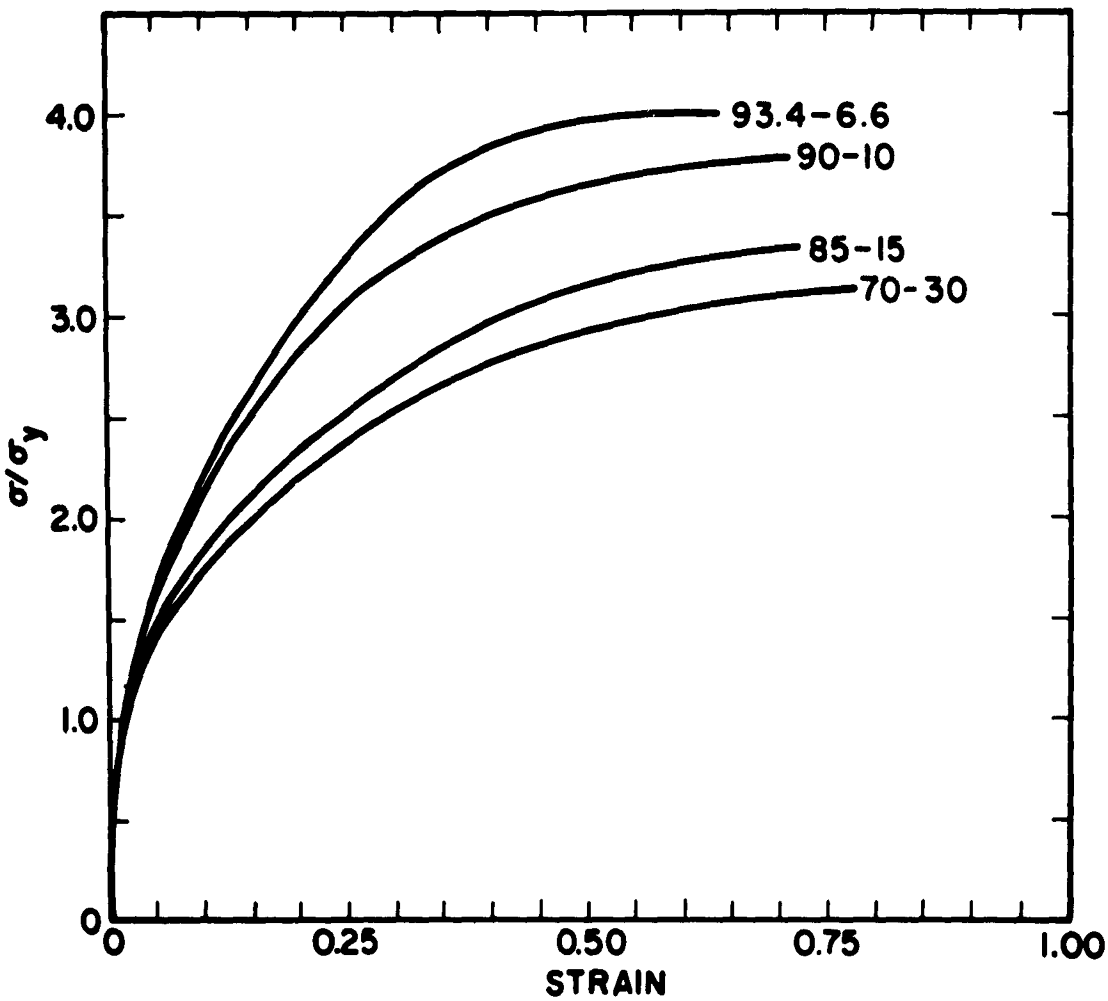

Figure 6. Normalized Tensile Behavior at 77.4°K

Thus, for a given value of σ/σ_y, the plastic strain during the first quarter cycle is less in the case 95-5 than in 70-30 α -brass. Moreover, the results of Avery and Backofen (ref 4) on the fatigue hardening of copper-aluminum alloys shows that the alloys with higher γ harden much more rapidly during cycling than do the low γ alloys. Therefore, for a given multiple of the yield stress, we can conclude that as cycling progresses, the higher zinc alloys will sustain a greater accumulated plastic strain than the lower zinc alloys. Thus, on the basis of a Coffin's law argument, one can conclude that the low zinc alloys will last longer in fatigue at the same value of σ/σ_y. Coffin's law is an empirical relation between plastic strain and cycles to failure. It has the form, $N^{\frac{1}{2}} \Delta\epsilon_p = C$, where N is the number of cycles, $\Delta \epsilon_p$ is the plastic strain range, and C is a constant related to the fracture ductility of the metal in question (ref 10).

The effect of stacking fault energy for lives $\leq 10^6$ cycles is therefore to be explained in terms of the deformation mechanisms which occur during cycling. For those metals which cross-slip easily, dislocation jog formation and dislocation loop formation can occur on a large scale and will account for the high rate of hardening and the ability to accommodate large multiples of the yield stress. In the low γ materials, slip is initially confined to a given glide plane and the work hardening rate is therefore lower. This type of behavior has been observed in α -brass single crystals (ref 11) and can be seen directly in the shear stress-strain curves. Thus it would appear that the effect of high stacking fault energy is to raise the work hardening rate, which allows the material to sustain a smaller plastic strain for a given value of the stress ratio, σ/σ_y. Some of the arguments given in the past (based on cross-slip) would therefore seem to be in error largely because of an unavoidable confusion which exists between the yield stress and γ.

In a further attempt to test the relationship between cross-slip and fatigue, the endurance limit (unreduced) was plotted against τ_{III}, the threshold stress for stage III hardening in single crystals of Cu-Zn (table 1). No meaningful relationship was apparent. However, a number of investigators have recently reported that cross-slip can occur in many materials before τ_{III} is reached. Moreover, all of the fatigue tests which we have conducted utilized stresses which would exceed τ_{III} in the most favorably oriented grains. However, from the requirements of constancy of volume and grain boundary continuity, any extensive plastic deformation in polycrystalline materials must be accompanied by deformation on several different slip systems (ref 12). Since τ_{III} is measured only in and only for single crystals deforming initially only on one slip system, the lack of a meaningful relationship mentioned earlier was neither surprising nor disquieting. It seems improbable that any meaningful correlation could ever be established between the endurance limit and τ_{III} in polycrystalline materials tested over a reasonable range of stresses.

To further aid the interpretation of data, microbeam, back-reflection X-ray photographs were obtained from fully fatigued samples of each alloy. The interpretation of such photographs has been discussed in a previous paper (ref 13). In all cases, the photographs showed extensive arcing of spots around the Debye-Scherrer ring which indicates a break-up of grains into subgrains. This break-up is much more severe adjacent to the fatigue failure in the case of the softer alloy (95-5). In no cases could the arcs be resolved into individual spots (X-ray beam divergence 1.5×10^{-4} radian), which indicates significant dislocation content within the subgrains. This observation is in agreement with previous results on the nature of substructure in α -brass (ref 14).

Finally, we have analyzed the data of Burghoff and Blank (ref 15) who obtained S-N curves for a similar range of Cu-Zn alloys at room temperature. Apparently, the effects of diffusion and other thermally activated processes at this temperature are sufficient to mask the effects of dislocation interaction because we were unable to obtain any consistent correlation with γ, τ_{III}, or with our data.

Because of the possibility that the yield stress itself may be significantly affected by the magnitude of γ in polycrystalline aggregates, the method of reduced stresses used here may be misleading. It is entirely possible, for example, that the ability to transmit stress across a grain boundary by means of dislocation pile-ups is a function of γ. Thus, in low γ materials where pile-ups are often observed at a grain boundary, plastic flow may occur at a lower total strain than for high stacking fault energy materials in which the ease of cross-slip will diminish the ability to transmit stress across a grain boundary. There can be little doubt that this is true at low strains; however, the 0.2 percent plastic strain accumulated in a test of yield strength may exceed the deformation range in which this effect can be observed. Nevertheless, the results shown in figure 5 may be somewhat misleading, because with the very artifice by which we eliminated the unwanted effects of yield strength we may have also eliminated the effect of stacking fault energy.

It is interesting to speculate on what may occur during long life fatigue ($>10^8$ ~). In figure 4(a) the variation of σ_e with γ at 10^6 cycles does tend in the direction which currently popular ideas concerning the roles of subgrain formation and subgrain boundaries would predict. If one is willing to ignore all of the other data, he can quickly construct a model which will explain, in a qualitative fashion, the observed behavior at 10^6 cycles. Ignoring the other data may be justifiable if one accepts the concept that there is a change in the dominating mechanism, as was mentioned earlier. Unfortunately we have not data for lives greater than 10^6 cycles. Such data might well confirm the importance of cross-slip in polygonization and crack growth, and it might give a clearer picture of the effect of stacking fault energy on long life endurance limit. But from our data, one can not predict with any certainty that this would be observed.

In order to develop a model which would predict, or even describe, the fatigue behavior of f.c.c. polycrystalline metals as a function of stacking fault energy, it would first be necessary to develop a model which could explain yielding and work hardening phenomena on the basis of stacking fault energy. This model would have to include the effects of the initial condition of the metal, strain rate, mean strain, strain amplitude, testing temperature, and a host of other variables too numerous to mention. In the event that this could be successfully accomplished, it might be possible to conduct an experiment in which enough of the variables were known, controlled, or measured to yield results which could be explained on the basis of stacking fault energy. In view of our limited knowledge of the various parameters and their interactions, and in view of the paucity of data on stacking fault energies, such an experiment does not presently seem feasible.

CONCLUSIONS

Fatigue endurance limit behavior seems to have a dual and opposite dependence on yield strength and stacking fault energy. For stresses high enough to produce a significant amount of work hardening, the endurance limit increases as the stacking fault energy increases. This is a manifestation of a higher work hardening rate for high stacking fault energy materials. In the long life region this trend is probably reversed. At the present time it is not possible to devise a simple model which will predict the observed results.

REFERENCES

1. Ham, R. K., Phil. Mag., V. 7, p. 95, 1962.
2. Segall, R. L., Partridge, P. G. and Hirsch, P., Phil. Mag. V. 6, p. 1493, 1961.
3. Feltham, P., Phil. Mag., V. 6, p. 1479, 1961.
4. Avery, D. H. and Backofen, W. A., "The Nucleation and Growth of Fatigue Cracks", Proceedings of the Lake Wilderness, AIME Fracture Conference, 1962, (in press).
5. Seeger, A., "Dislocations and Mechanical Properties of Crystals", p.243, Wiley and Sons, 1957.
6. Howie, A. and Swann, P., Phil. Mag., V. 6, p. 1215, 1961.
7. Thornton, Mitchell, and Hirsch, Phil. Mag., V. 7, p. 1349, 1962.
8. Allen, N. T., Scofield, T. H. and Tate, A. E. L., Nature, V. 168, p. 378, 1951.
9. Hibbard, W. R., Jr., Trans. Met. Soc., AIME, V. 212, p. 1, 1958.
10. Tavernelli, J. F. and Coffin, L. E., Jr., Trans. ASM, V. 51, p. 438, 1959.
11. von Goler, K. F. and Sachs, G., Z. Physik, <u>55</u>, p. 581, 1929.
12. Taylor, G. I., J. Inst. Metals, V. 62, p. 307, 1938.
13. Grosskreutz, J. C., Journal of Applied Physics, 1962, (in press).
14. Swann, P. R. and Nutting, J., Journal of the Institute of Metals, V. 90, p. 133, 1961.
15. Burghoff, H. L. and Blank, A. I., Proc. ASTM 48, p. 709, 1948.

APPENDIX

Individual Curves and Numerical Data

TABLE 2

NUMERICAL DATA - STRESS VS CYCLES TO FAILURE

ALLOY	STRESS	LIFE	ALLOY	STRESS	LIFE
93.4-6.6	$\times 10^{-4}$	$\times 10^{-4}$	85-15	$\times 10^{-4}$	$\times 10^{-4}$
	2.28	33.37		2.43	26.35
	2.88	0.49		*2.10	10.53
	*2.60	0.55		*2.14	184.97
	2.19	9.31		2.34	35.57
	2.57	3.13		2.64	2.30
	3.50	0.07		2.80	0.37
	2.45	3.60		2.91	1.14
	1.84	59.16		3.42	0.06
	3.36	0.57		2.06	22.06
	2.00	15.42		1.94	412.37
	2.06	15.09		1.98	282.56
	1.88	47.39		2.22	10.43
	1.71	122.85			
	*2.60	3.73			
90-10	2.30	6.35	70-30	2.16	111.72
	2.07	34.45		2.36	23.17
	1.92	67.35		2.33	21.71
	3.12	0.34		*2.52	5.38
	2.62	7.23		*2.54	16.20
	1.98	158.95		2.84	0.41
	*2.21	38.99		2.23	55.55
	2.36	13.50		2.69	3.35
	1.98	146.13		2.42	6.93
	1.98	27.81		2.39	14.46
	2.46	9.51			
	*2.23	6.14			
	2.54	10.09			
	2.82	1.19			

* Plotted as average

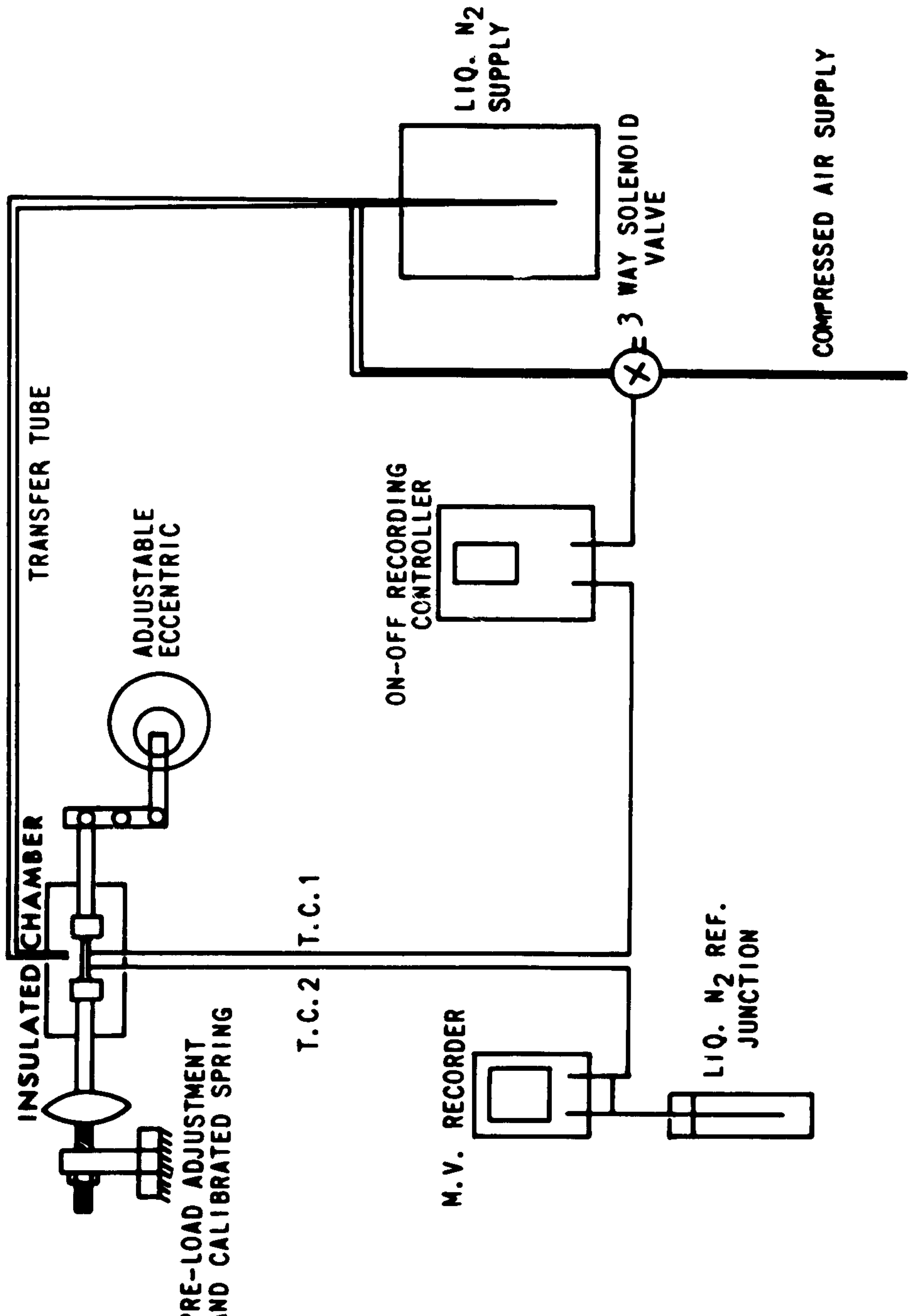

Figure 7. Specimen Installation and Instrumentation

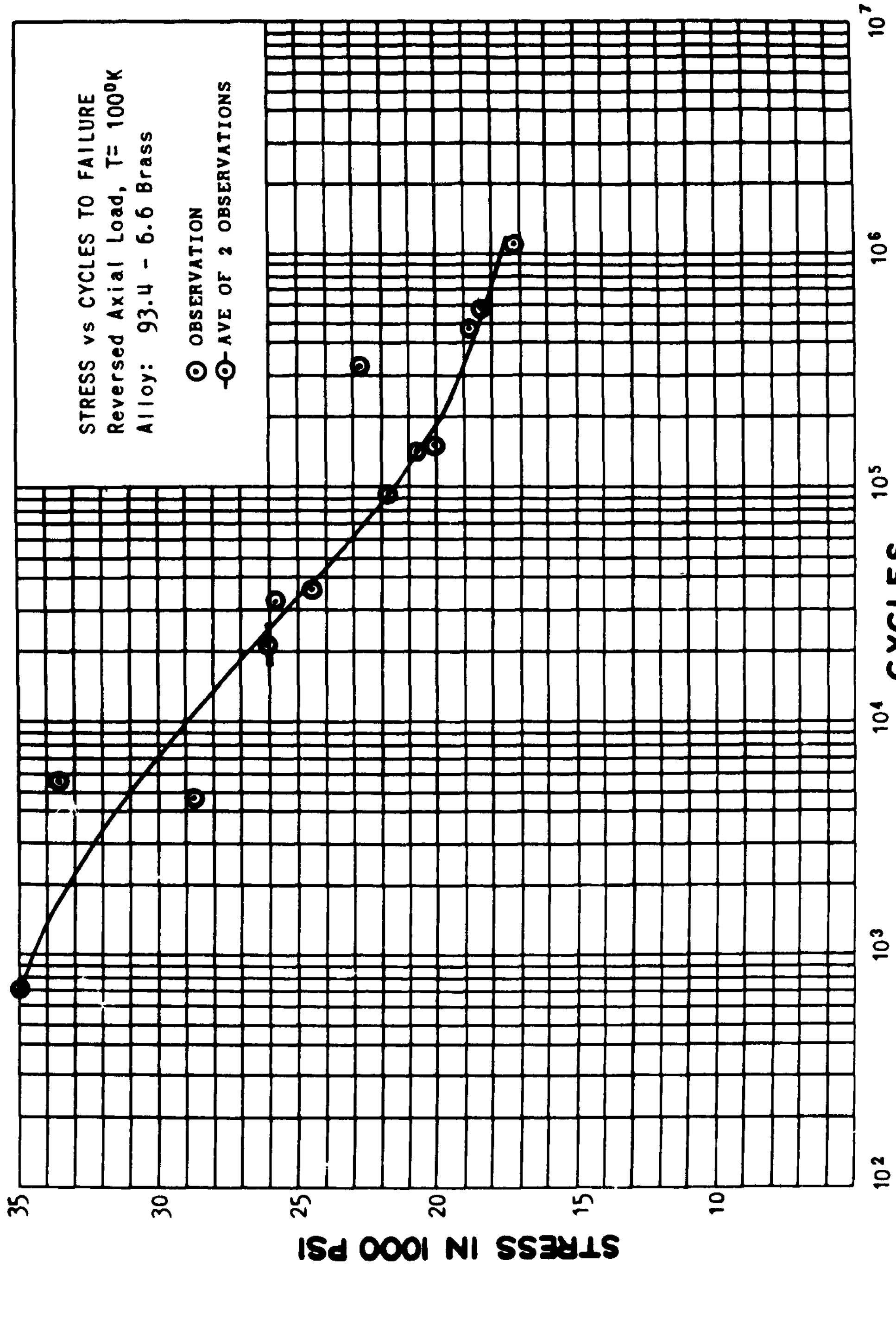

Figure 8. Stress versus Cycles to Failure, Alloy 93.4-6.6 Brass

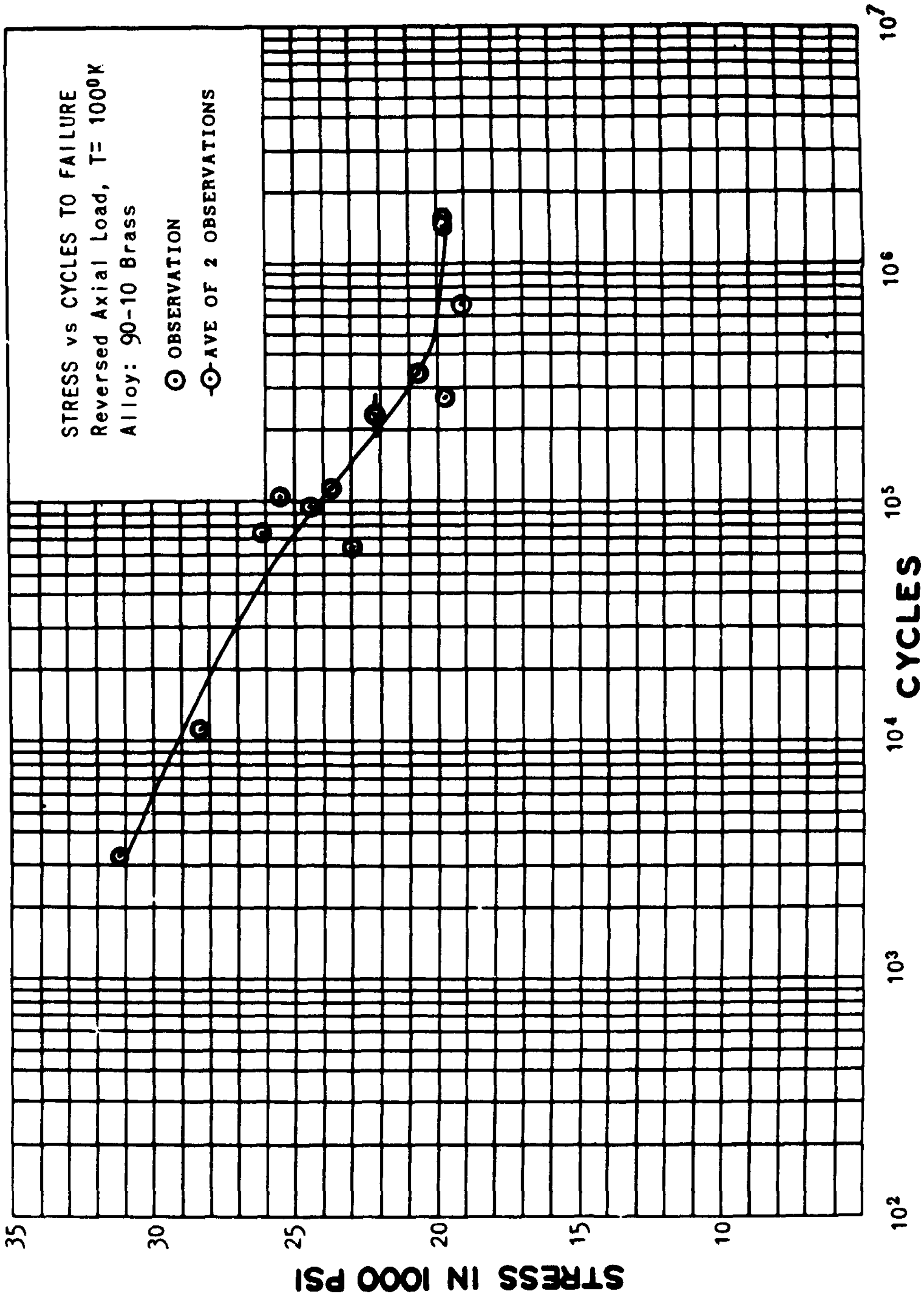

Figure 9. Stress versus Cycles to Failure, Alloy 90-10 Brass

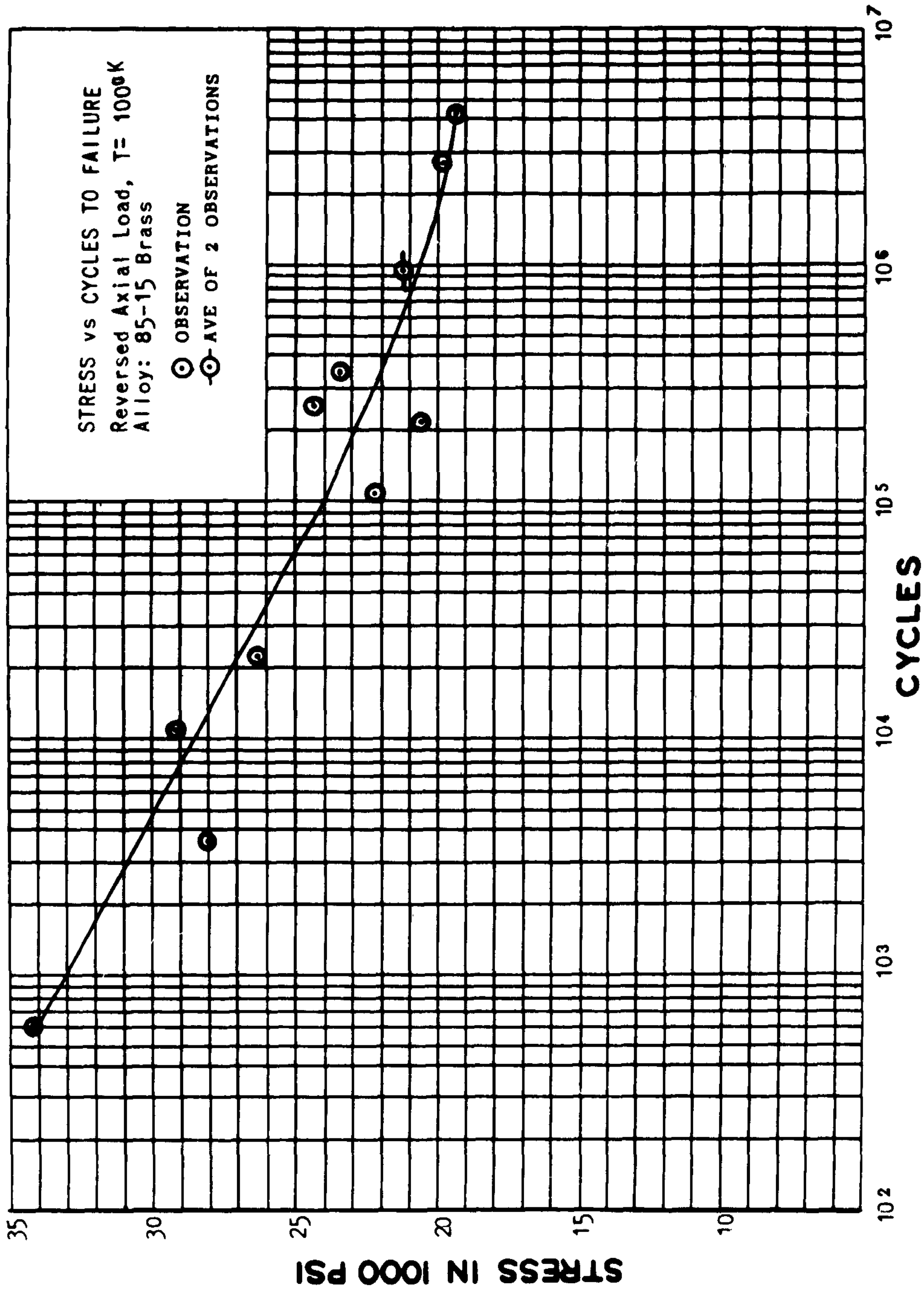

Figure 10. Stress versus Cycles to Failure, Alloy 85-15 Brass

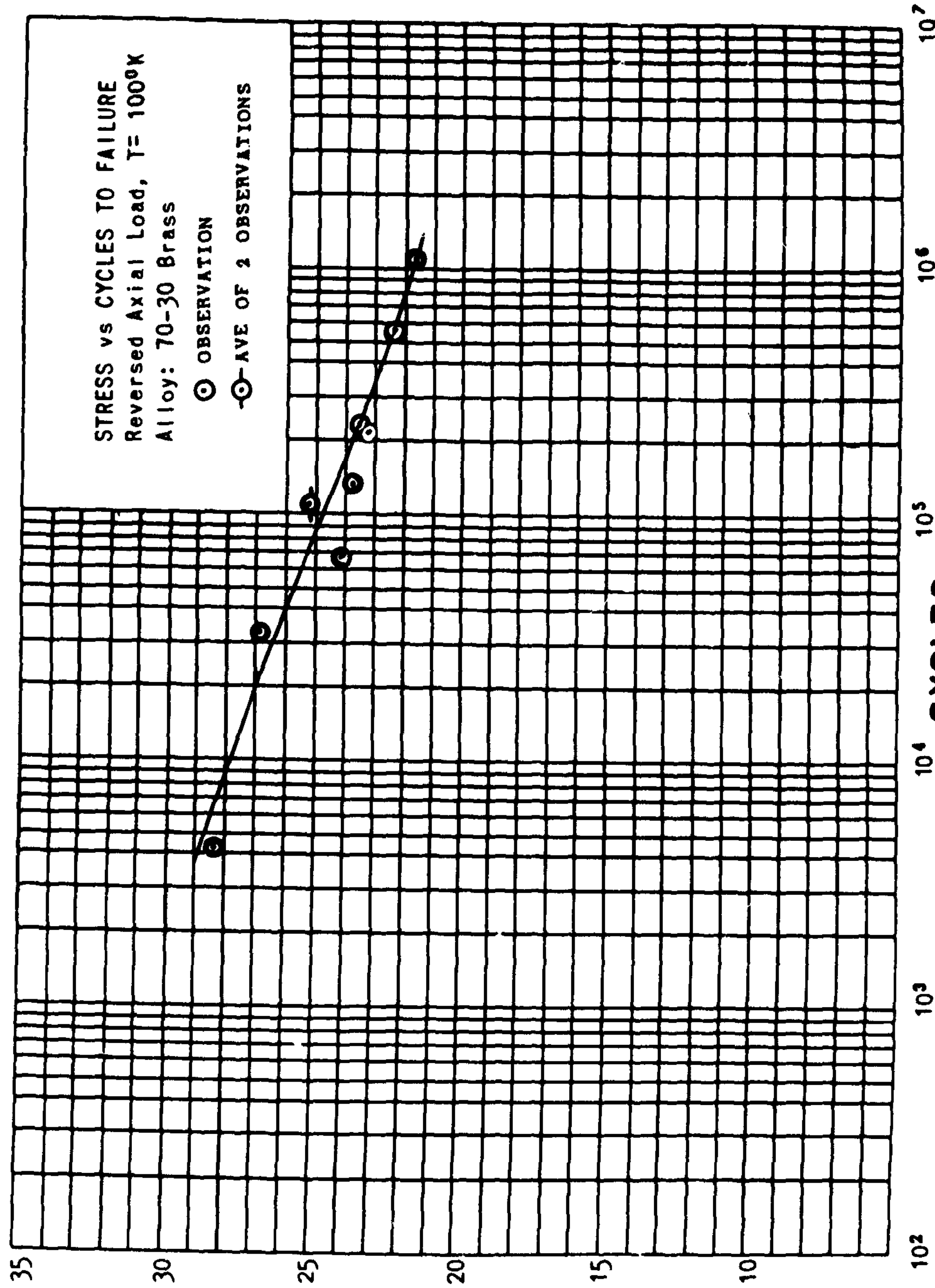

Figure 11. Stress versus Cycles to Failure, Alloy 70-30 Brass

Aeronautical Systems Division, Dir/Materials and Processes, Metals & Ceramics Lab, Wright-Patterson AFB, Ohio.
Rpt Nr ASD-TDR-63-252. FATIGUE OF COPPER-ZINC ALLOYS AT 100°K. Final report, Apr 63, 21p. incl illus., tables, 15 refs.
Unclassified Report

Fatigue tests were conducted to determine the S-N diagrams for a series of copper-zinc alloys at 100°K. An attempt was made to relate the endurance limit behavior to both stacking fault energy and yield strength, but no simple relationship was found. It is suggested that high stacking fault energy increases the cyclic work hardening rate by

(over)

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1. Fatigue tests
2. Copper-zinc alloys
I. AFSC Project 7351 Task 735106
II. J. A. Roberson
J. C. Grosskreutz
III. Aval fr OTS
IV. In ASTIA collection

Aeronautical Systems Division, Dir/Materials and Processes, Metals & Ceramics Lab, Wright-Patterson AFB, Ohio.
Rpt Nr ASD-TDR-63-252. FATIGUE OF COPPER-ZINC ALLOYS AT 100°K. Final report, Apr 63, 21p. incl illus., tables, 15 refs.
Unclassified Report

Fatigue tests were conducted to determine the S-N diagrams for a series of copper-zinc alloys at 100°K. An attempt was made to relate the endurance limit behavior to both stacking fault energy and yield strength, but no simple relationship was found. It is suggested that high stacking fault energy increases the cyclic work hardening rate by

(over)

increasing the probability of dislocation intersection and jog formation. The increase in work hardening rate is reflected in a decrease in plastic strain amplitude, and a subsequent increase in fatigue life. These arguments are bounded on one side by considerations of the yield strength of the alloys, and on the other side by consideration of the dominant mechanisms operative in short and long life fatigue.

1. Fatigue tests
2. Copper-zinc alloys
I. AFSC Project 7351 Task 735106
II. J. A. Roberson
J. C. Grosskreutz
III. Aval fr OTS
IV. In ASTIA collection